MINING PETROLEUM

FUNDAMENTAL STUDIES RELATED TO THE ORIGIN AND NATURE OF CREEP OF METALS

TENTH TECHNICAL REPORT

STRAIN HARDENING OF LATENT SLIP SYSTEM IN ZINC CRYSTALS

BY

E. H. EDWARDS
J. WASHBURN
E. R. PARKER

APRIL 195[illegible]

SERIES NO. 27 ISSUE NO. 10
CONTRACT N7-ONR-29516, NR 039-009

MINERALS RESEARCH LABORATORY
INSTITUTE OF ENGINEERING RESEARCH
UNIVERSITY OF CALIFORNIA, BERKELEY

FUNDAMENTAL STUDIES RELATED TO THE ORIGIN AND NATURE OF CREEP OF METALS

Tenth Technical Report

Strain Hardening of Latent Slip Systems in Zinc Crystals

By

Eugene H. Edwards
Jack Washburn
Earl R. Parker

Office of Naval Research Project

N7-onr-29516 NR 039-009

April, 1954

University of California
Berkeley, California

Report No. 10
Series No. 27

INTRODUCTION

Strain hardening which accompanies slip in a metal crystal is not limited to the slip systems actively contributing to the plastic strain. There is also a strengthening of the latent slip systems which are crystallographically equivalent to the active system. In the case of aluminum deformed in tension, Taylor and Elam(1) found that plastic flow on one set of octahedral planes caused either the same or a slightly greater hardening on an inactive set. Similar results have been obtained for crystals of α brass(2) and of copper-aluminum solid solutions(3). Experiments by Röhm and Kochendörfer(4) indicated that the hardening of the active system exceeded the hardening of any latent system for aluminum crystals deformed in shear. On the other hand strain hardening in zinc and cadmium crystals tested in simple shear at -196° C was recently found to be greater in the latent than in the active slip direction(5). When the strain direction was shifted during testing to a direction 60° from the original, a higher stress was required for glide to continue than would have been needed for flow to proceed in the original direction.

The relation between the hardening of active and latent systems must be complex, depending upon the relative orientations of the systems and upon the experimental techniques used to deform the crystal. A detailed knowledge of the hardening produced in latent systems would be valuable in choosing between possible dislocation models of the strain hardening process. The experimental data now available are incomplete and even contradictory. The present study was undertaken to provide quantitative information regarding the relative amount of hardening on active and latent systems for a particularly simple case. Zinc crystals were deformed in simple shear along one of the three crystallographically equivalent directions $[2\bar{1}\bar{1}0]$, $[\bar{1}2\bar{1}0]$ and $[\bar{1}\bar{1}20]$ in the slip plane and the relative hardening produced in each of these directions was compared.

EXPERIMENTAL PROCEDURE AND RESULTS

The single crystals of 99.99 percent purity zinc used in this investigation were grown from the melt in the form of one-inch diameter spheres. Shear specimens were acid machined from the crystals by a method which has been described previously(6). The gage section of the test specimen was a cylinder having a height of 1/8 inch and a cross-sectional area of approximately 1/3 square inch. All specimens were machined with the axis of this cylindrical gage section aligned with the [0001] axis of the crystal.

An innovation in the design of the shearing apparatus for these tests was the provision for a rapid shift in the direction of the applied stress from one slip direction to any other slip direction lying in the same slip plane. Fig. 1 shows a section through the specimen and shearing apparatus, and Fig. 2 is a photograph of the complete assembly with extensometer in position. During a test the specimen and shearing apparatus were submerged in liquid nitrogen. A shift from one slip direction to another could be accomplished without removal of the assembly from the nitrogen, thus avoiding the possibility of recovery occurring during the interval in which the direction change was being made.

The observation that hardening of a latent slip system may exceed that of the active system appears to be generally valid for zinc. A considerable number of crystals with somewhat different previous histories with regard to

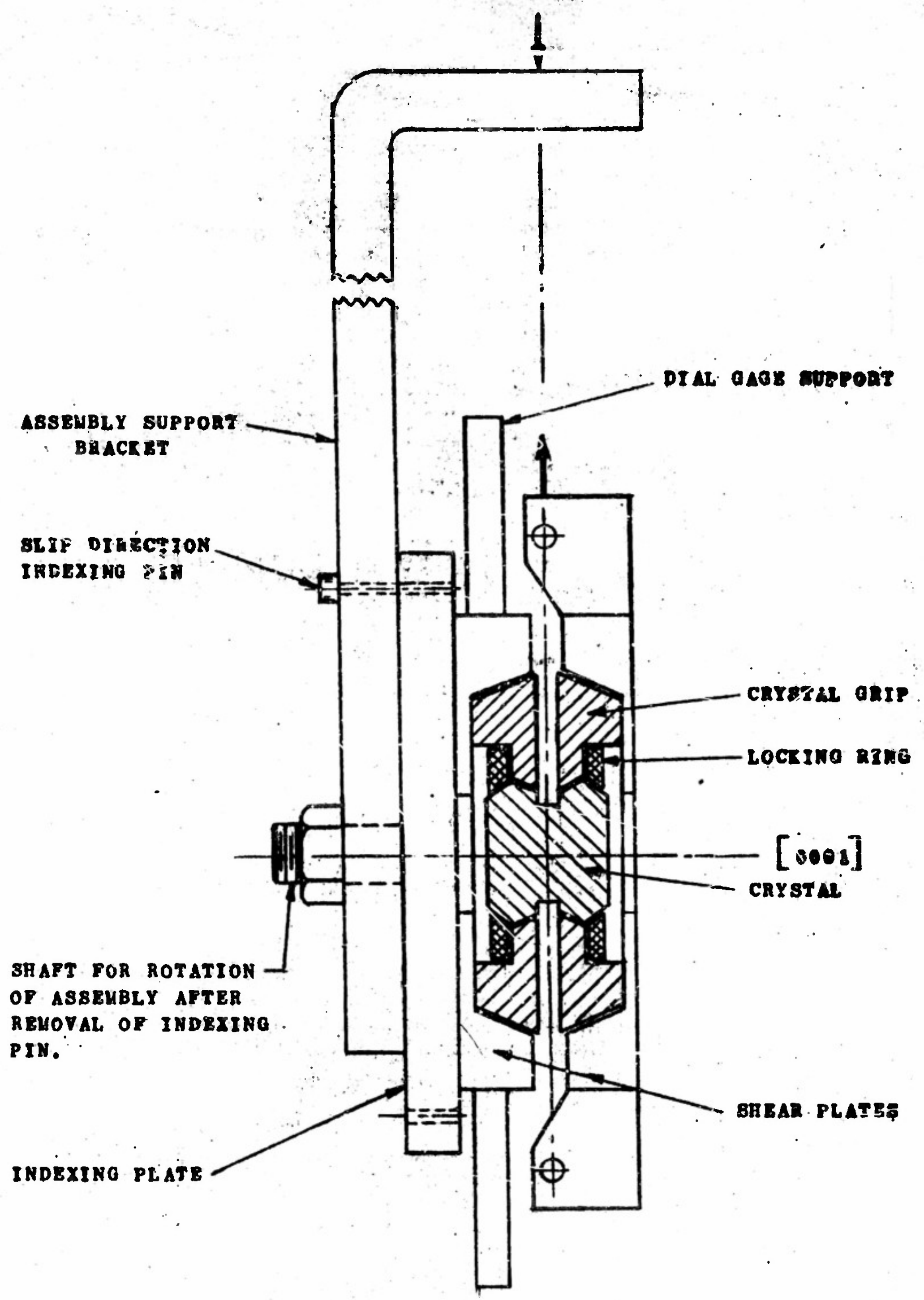

FIG. 1 SCHEMATIC DIAGRAM OF APPARATUS USED FOR TESTING CRYSTALS IN SIMPLE SHEAR AT -196°C. DIRECTION OF STRAINING COULD BE SHIFTED DURING TESTING FROM A GIVEN SLIP DIRECTION TO ANY OTHER SLIP DIRECTION LYING IN THE SAME SLIP PLANE. ASSEMBLY WAS SUBMERGED DURING TESTING IN LIQUID NITROGEN CONTAINED IN A DEWAR FLASK.

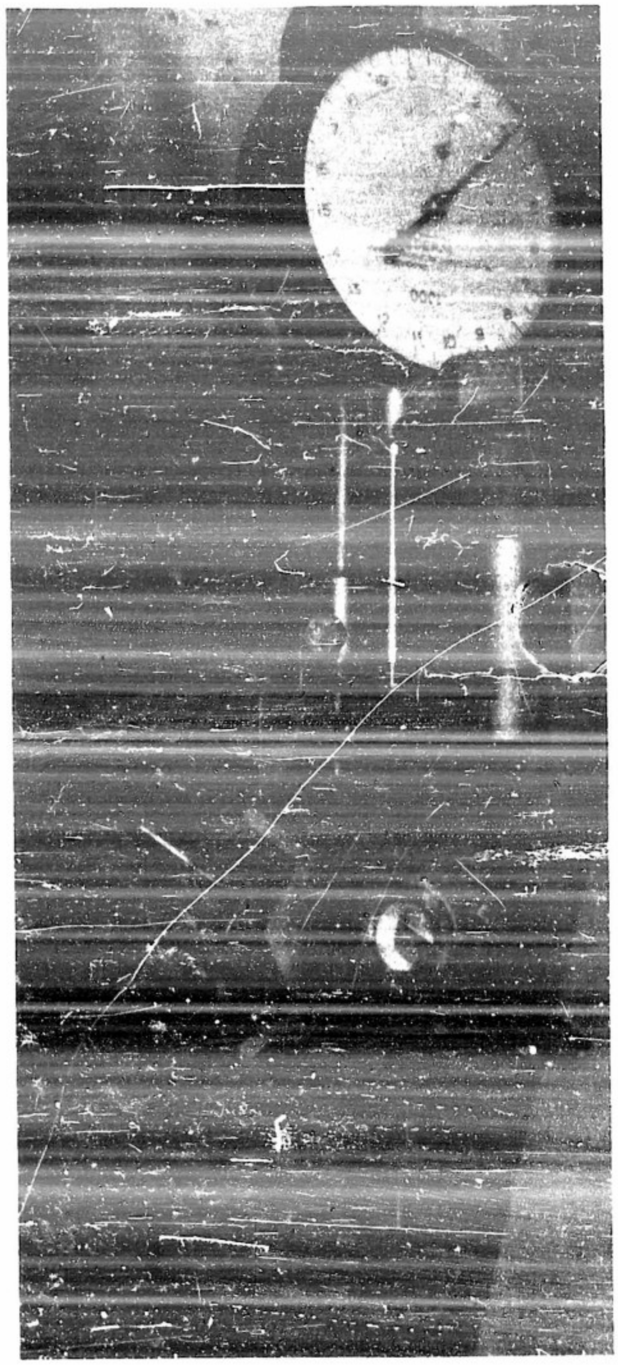

FIG. 2 ROTATION ASSEMBLY SHOWING A SHIFT IN THE DIRECTION OF STRAINING FROM ONE SLIP DIRECTION TO A NEW DIRECTION 120° FROM THE ORIGINAL IN THE SAME CRYSTALLOGRAPHIC SLIP PLANE.

orientation during growth, heat treatment, and macro-substructure were strained approximately 4 percent in a given slip direction at -196° C, unloaded, and immediately strained in a second slip direction 60° from the original direction. In every instance the shear stress required to cause slip to begin in the new direction was considerably higher than the stress which would have been necessary for slip to continue in the original direction.

The strain hardening, including the increment due to this change in slip directions, could be completely eliminated by annealing. Fig. 3 shows three successive stress-strain curves obtained from the same crystal for which the direction of straining was shifted during each test to a new slip direction 60° from the first after a strain of 2.6% in the original direction. The crystal was annealed for one hour at 400° C before each test. Selection of any one of the three slip directions as the first active direction after annealing produced the same set of curves showing that the crystal had been returned to its original isotropic condition by the annealing treatment. Because of the complete recovery, it was practicable to study in the same crystal the magnitude of the hardening of each latent system in relation to the amount of previous strain in the active system. Errors which might have been introduced by comparing data obtained from crystals of different degrees of internal perfection were thus avoided.

The shear stress increase required to cause plastic flow to take place in the second slip direction was found to increase as a function of strain in the original direction. A series of tests was performed on a crystal in which the strain in the original direction was varied from 0.3 to 10 percent. Representative examples of the individual tests are reproduced in Figs. 4 and 5, showing the effect of shifting to a new slip direction 60° and 120°,

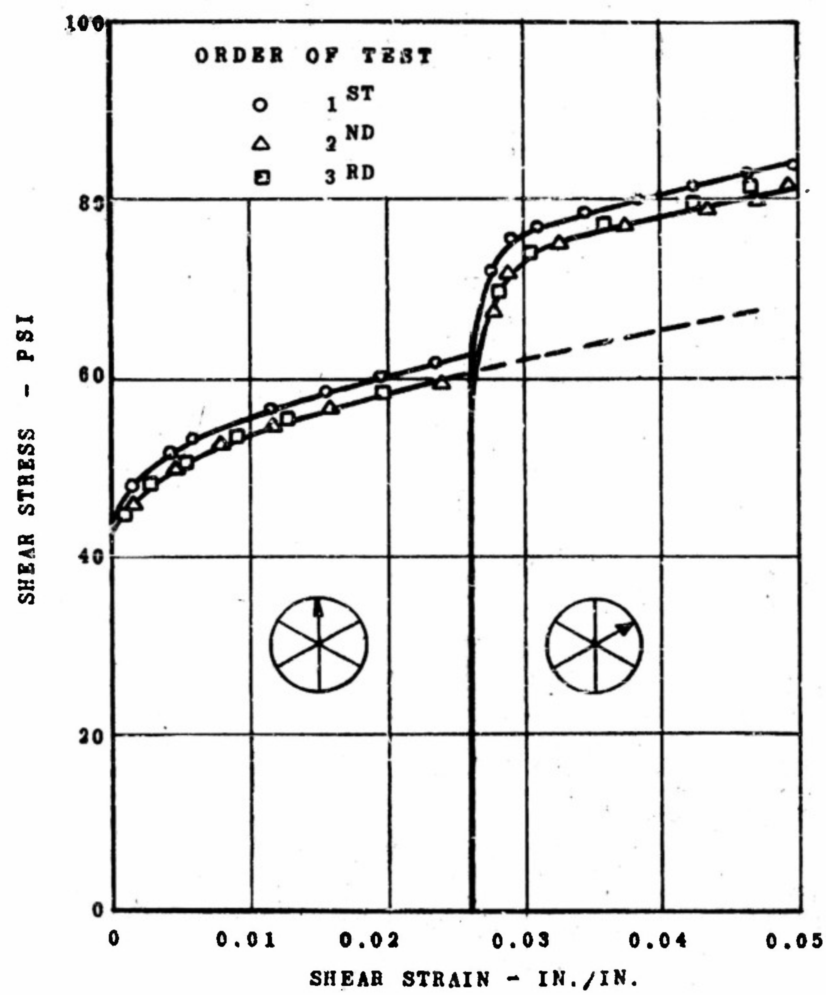

FIG. 3 THREE SUCCESSIVE STRESS-STRAIN CURVES FROM A ZINC CRYSTAL SHOWING THE EFFECT OF SHIFTING THE STRAIN DIRECTION DURING TESTING TO A NEW DIRECTION 60° FROM THE ORIGINAL DIRECTION OF SLIP. COMPLETE RECOVERY OF MECHANICAL PROPERTIES WAS ACHIEVED BY AN ANNEAL OF ONE HOUR AT 400°C BEFORE EACH TEST.

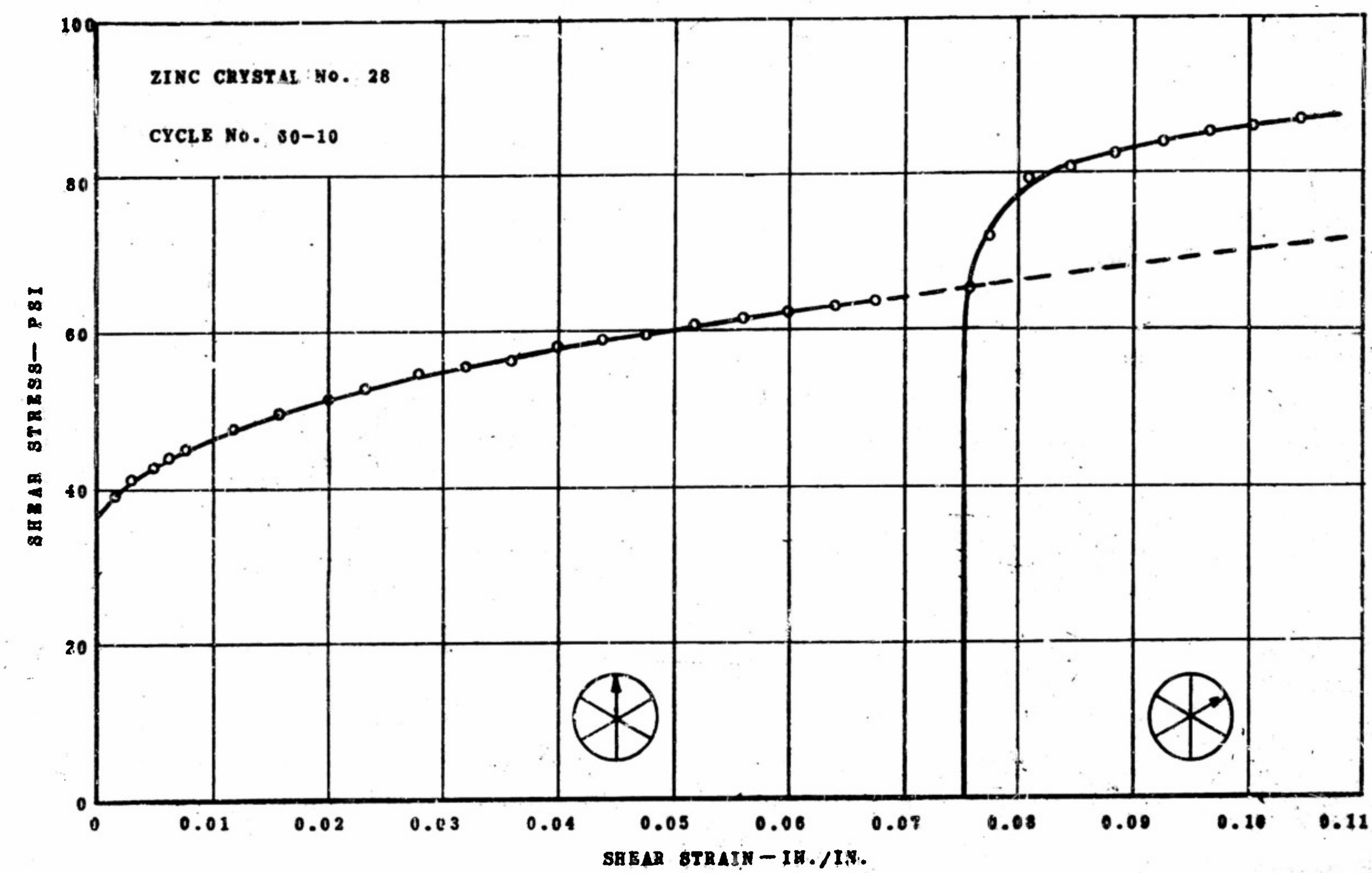

FIG. 4 STRESS-STRAIN CURVE FOR A ZINC CRYSTAL SHOWING EFFECT OF SHIFTING THE STRAIN DIRECTION DURING TESTING AT −196°C TO A NEW DIRECTION 60° FROM THE ORIGINAL DIRECTION OF SLIP AFTER STRAIN OF 0.075 IN THE ORIGINAL DIRECTION.

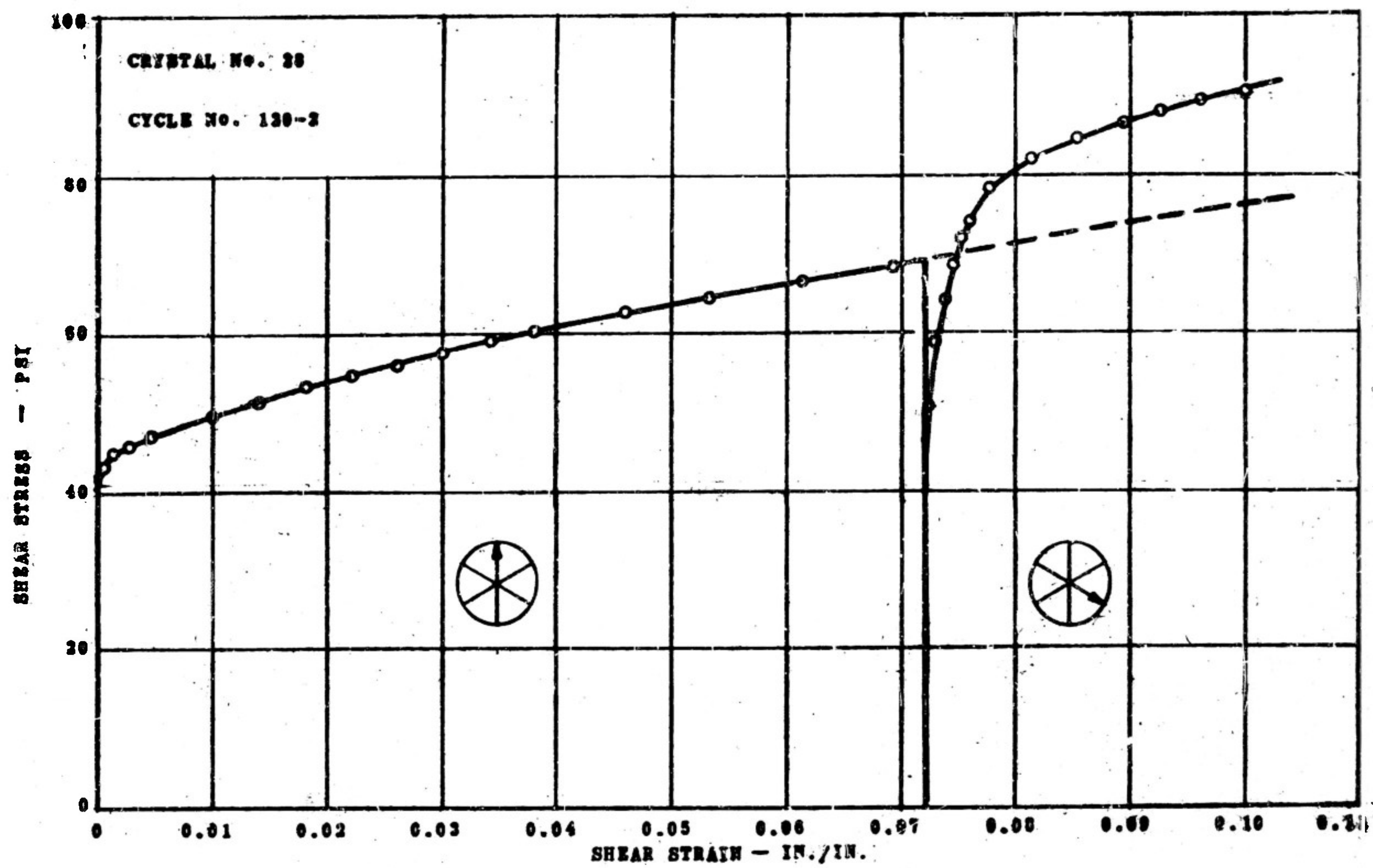

FIG. 5 STRESS-STRAIN CURVE FOR A ZINC CRYSTAL SHOWING EFFECT OF SHIFTING THE STRAIN DIRECTION DURING TESTING AT −196 °C TO A NEW DIRECTION 120° FROM THE ORIGINAL DIRECTION OF SLIP AFTER STRAIN OF 0.072 IN THE ORIGINAL DIRECTION.

respectively, from the original direction. Fig. 6 is a plot obtained from these data indicating that the stress increase required for flow to begin in the second direction rises sharply for small strains in the original direction. At larger strains the rate of increase is less rapid and becomes linear with strain in the first direction. All the data recorded in Fig. 6 were obtained from a single specimen. The only significant difference between the results obtained from 120° shifts of direction as compared to those from 60° shifts was the occurrence of a small amount of strain at low stress levels when the strain direction was shifted 120°.

Another series of tests were performed in which the strain direction was changed by 180°, using the same crystal from which the data on 60° and 120° shifts in strain direction were obtained. A representative example of one of these tests is shown in Fig. 7. A measureable amount of strain in the reverse direction was always observed at a very low applied stress and a considerable reverse strain occurred before a stress equal to the maximum achieved in the forward direction was reached. Even at very large strains in the reverse direction, the stress was less than the value it would have reached at the same total strain had the reversal not taken place. A portion of the strain hardening was thus permanently lost as a result of the reversal. Fig. 8 shows that the stress-strain curve after reversal of the strain direction fell below the extrapolated original curve by an amount that increased linearly with strain in the original direction. The decrease in the flow stress plotted in Fig. 8 was measured after a strain of 3 percent in the reverse direction.

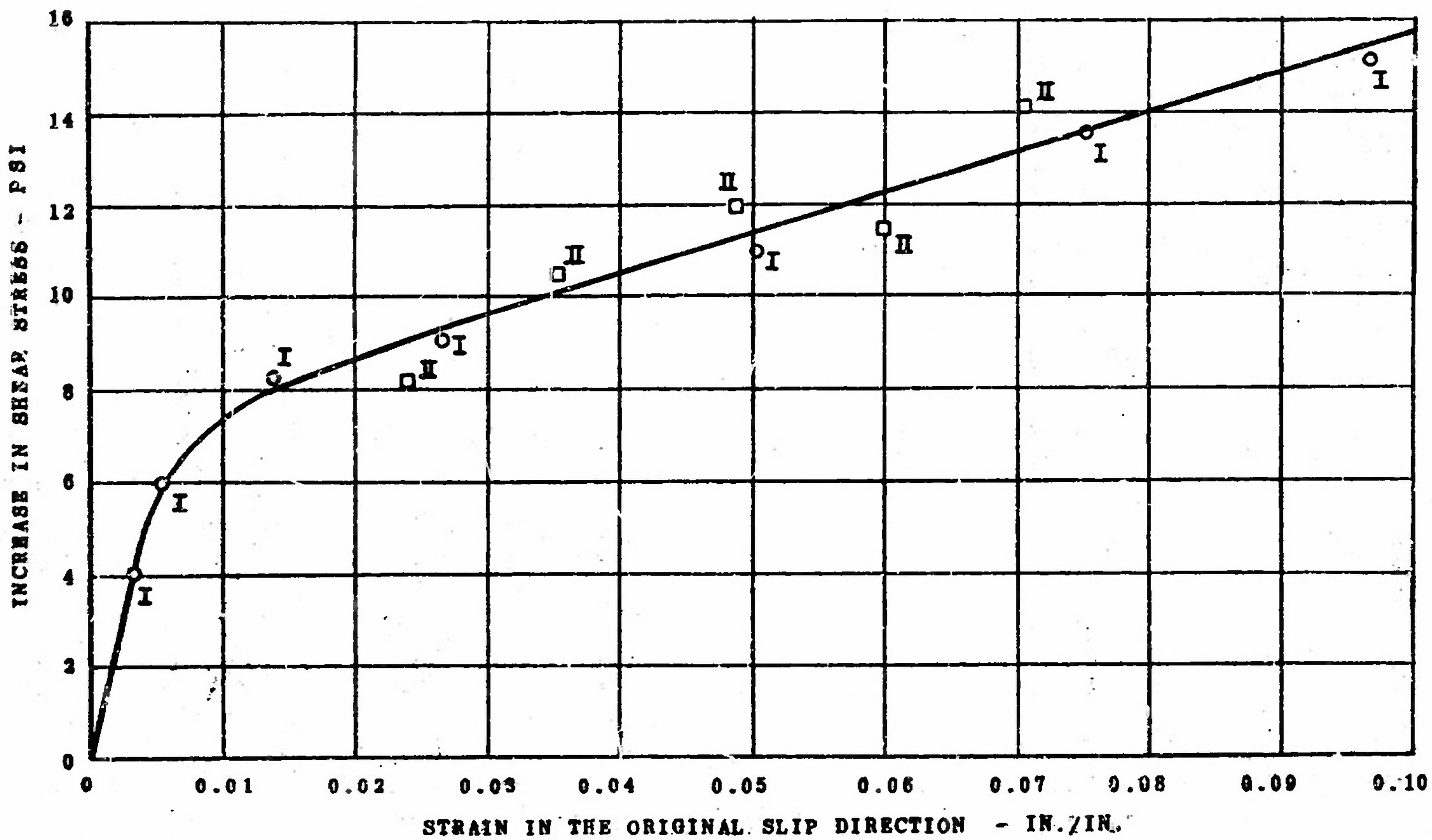

FIG. 6 INCREASE IN SHEAR STRESS AS A RESULT OF SHIFTING THE STRAIN DIRECTION TO A NEW SLIP DIRECTION DURING TESTING AT −196°C AS A FUNCTION OF STRAIN IN THE ORIGINAL SLIP DIRECTION:

I NEW SLIP DIRECTION 60° FROM ORIGINAL DIRECTION
II NEW SLIP DIRECTION 120° FROM ORIGINAL DIRECTION

STRESS INCREASE MEASURED AFTER A STRAIN OF 0.03 IN THE NEW DIRECTION.

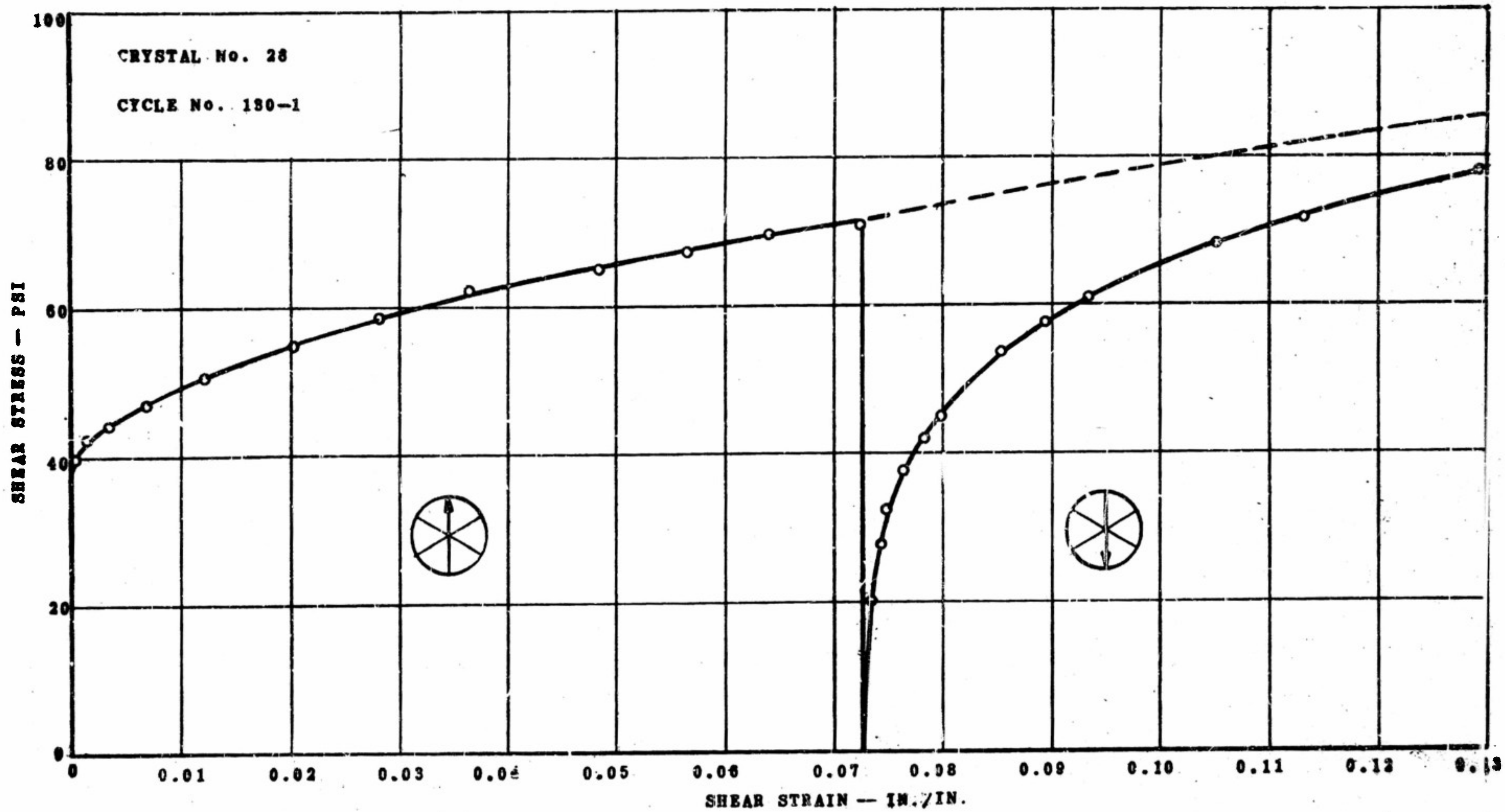

FIG. 7 STRESS-STRAIN CURVE FOR A ZINC CRYSTAL SHOWING EFFECT OF REVERSING THE DIRECTION OF STRAINING AFTER A STRAIN OF 0.073 IN THE ORIGINAL DIRECTION DURING TESTING IN SIMPLE IN SIMPLE SHEAR AT —196 °C.

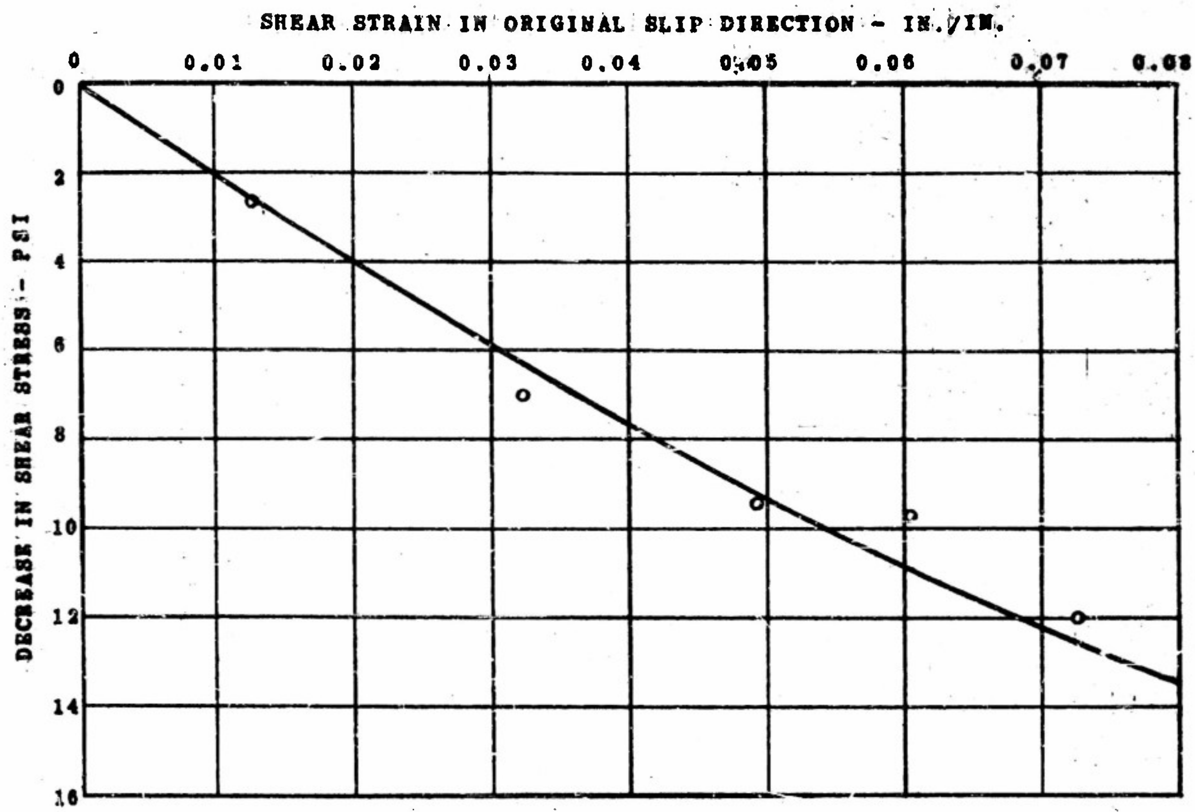

FIG. 8 CURVE SHOWING DECREASE IN SHEAR STRESS ACCOMPANYING REVERSAL OF DIRECTION OF STRAINING DURING TESTING IN SIMPLE SHEAR AT −196°C AS A FUNCTION OF THE STRAIN IN THE ORIGINAL SLIP DIRECTION. DECREASE IN STRESS MEASURED AFTER 0.03 STRAIN IN THE REVERSE DIRECTION.

DISCUSSIONS AND CONCLUSIONS

No adequate theory for the strain hardening accompanying the plastic deformation of metal crystals by simple slip (easy glide, lamelar slip) has yet been proposed. Kochendörfer(4) has suggested that this hardening, which is approximately linear with strain, is due to the action of back stresses created by dislocations piled up at regions of crystal imperfection. The local stress at dislocation sources was assumed thereby to be decreased and continued slip would then require an increase in applied stress. Mott(7) has suggested that the primary barrier to the motion of dislocations during simple slip is probably other dislocations which cut across the active slip plane and have a Burgers vector component normal to the slip plane. In cubic structures these dislocations could lie on other equivalent slip planes; in hexagonal crystals they would be dislocations with a Burgers vector in the c direction. The motion of the screw segments of dislocation loops expanding from a source in the slip plane would be retarded by these intersecting dislocations. Read(8) has pointed out that the crossing of two screw dislocations at right angles produces a step in each of the dislocations. Any motion of the stepped screw dislocation is accompanied by the formation of a row of lattice vacancies or interstitial atoms which are left behind the moving step.

The experimental results obtained from the 180° reversal of applied stress are not inconsistent with the thought that a strained crystal may contain dislocation loops which have encountered obstacles and therefore have not been able to spread completely across the slip plane. The low stress required to initiate strain in the reverse direction appears to confirm the idea of a back stress due to such partially expanded loops. The linear manner in which the permanent softening due to a reversal increases with strain in the first direction suggests that the number of dislocations held up behind obstacles increases linearly with strain and that the linear strain hardening is associated with the presence of these immobilized dislocations.

The results of 60° and 120° shifts of strain direction are not, however, easily understood on the basis of the simple picture presented above because it leads to the conclusion that the strain hardening should be greatest in the initial direction. The component of back stress acting in any direction in the same slip plane should vary with the cosine of the angle between this direction and the initial direction. This conclusion was pointed out by Kochendörfer(4) and apparently was confirmed by his results obtained on aluminum crystals using an experimental technique somewhat less direct than that employed here.

The present results obtained for hexagonal crystals in which only one slip plane exists were not consistent with this conclusion. The hardening in 60° and 120° directions was greater than the hardening in the initial direction. Only a component of the back stress due to immobile dislocations left by the strain in the first direction would act in the 60° and 120° directions. In the 120° direction this component of back stress would add to the applied stress. Therefore the action of back stresses on the sources of dislocations does not appear to be the controlling factor. An alternative approach is to assume that the effectiveness of barriers increases with increasing strain; a fraction of the dislocations moving through the crystal becoming arranged into metastable arrays which in turn tend to trap additional moving dislocations. The simplest example of such a metastable array is a wall of like edge dislocations in a plane at right angles to their common Burgers vector. The magnitude of the back stress is then dependent upon the strength of the barrier. Strain can only proceed when a stress great enough to move dislocations through the obstacles is reached. The low stress required for initial strain in the back direction would then be due to backward motion of dislocations piled up against the barriers but not incorporated into metastable networks. The permanent loss of strain hardening accompanying continued strain in the back direction might be due to dislocations of opposite sign issuing from the same sources which were responsible for the strain in the first direction. These could have the effect of partially eliminating the barriers by combinations of dislocations of opposite sign. Only a shift of 180° in the direction of straining would be expected to be accompanied by a permanent loss of a portion of the strain hardening.

Dislocation reactions of the type $\frac{a}{3}[2\bar{1}\bar{1}0] + \frac{a}{3}[\bar{1}\bar{1}20] \rightarrow \frac{a}{3}[1\bar{2}10]$ between the dislocations being generated by the applied stress in the new direction and those already present due to strain in the first direction might create immobile nodes, resulting in the formation of a large number of barriers in the early stages of strain in the new direction. The magnitude of the hardening effect due to the barriers of this type might be expected to be proportional to the number of them formed, which might in turn be proportional to the density of dislocations left in the crystal after straining in the first direction.

Since the hardening due to simple unidirectional slip, as well as the additional increment of hardening produced by shifting to a new slip direction during a test can be completely removed by heating the crystal near the melting point, the barriers which are responsible for the hardening must be of a form which can disappear when dislocation climb is permitted. A dislocation arrangement built up from complete dislocation loops would be expected to have this property since it would be a network containing an equal number of interconnected positive and negative dislocation elements.

SUMMARY

When a zinc crystal is deformed in simple shear, anisotropic strain hardening occurs in which the inactive slip systems are hardened more than the active one. The hardening effect is substantially the same for latent systems whose directions lie at 60° and 120° to the slip direction of the active system. The hardening of the latent systems is possibly due to dislocation barriers formed in the crystal during the initial stages of plastic flow in the latent directions. These barriers to dislocation movement in latent directions are a type which can be removed by annealing.

When the direction of straining is reversed, plastic flow begins at a stress much lower than that required for the onset of slip in the original direction. This effect is attributed to movement of dislocations which are trapped at internal barriers during deformation in the forward direction but which can move freely in the reverse direction. The permanent loss of a portion of the strain hardening when the direction of straining is reversed may be due to annihilation of some of the dislocations trapped in the crystal during the first strain.

ACKNOWLEDGMENT

The authors are indebted to Dr. Betsy Ancker and C. H. Li for their helpful discussions, and to Leslie Seaborn and Robert Arneson for their assistance in the construction of apparatus.

REFERENCES

1. G. I. Taylor and C. F. Elam, Proc. Roy. Soc. A, 108, (1925), 28.

2. V. Goler and G. Sacks, Z. Physik, 55, (1929), 581.

3. C. F. Elam, Proc. Roy. Soc. A, 115, (1927), 133.

4. F. Röhm and A. Kochendörfer, Z. Metallkde, 41, (1950), 265.

5. E. H. Edwards, Jack Washburn, and Earl R. Parker, Trans. AIME, TP 3645E; Jour. Metals, Nov. 1953, O. 1525.

6. E. R. Parker and Jack Washburn, "Deformation of Single Crystals," ASM Seminar on Modern Research Techniques, 1952.

7. N. F. Mott, Phil. Mag., 43 (1952), 1151.

8. W. T. Read, "Dislocations in Crystals", (McGraw-Hill Book Co., 1953).

TECHNICAL REPORT
Distribution List
N7onr-29516 NR-039-009

	No. of Copies
Chief of Naval Research, Attn: Metallurgy Branch, Code 423, Office of Naval Research, Washington 25, D.C.	2
Director, Naval Research, Attn: Lab. Technical Information Officer, Washington 25, D.C.	6
Bureau of Aeronautics, USN, Washington, D.C.	
N.E. Promisel AE-41	3
Tech. Library TD-41	1
Bureau of Ordnance, Navy Dept., Washington, D.C.	
ReX	3
Tech. Library Ad-3	1
Naval Ordnance Lab., Washington, Materials Lab.	1
U.S. Bureau of Ships, Navy Dept., Washington, D.C.	
Code 330	3
Code 337L, Tech. Lib.	1
U.S. Naval Engineering Exp. Sta., Annapolis, Md., Metals Lab.	1
N.Y. Naval Shipyard, Director Material Lab., Brooklyn, Code 907	1
Post Graduate School, U.S. Naval Academy, Monterey, Calif., Attn: Metallurgy Dept.	1
Supt., Naval Gun Factory, Washington, D.C.	1
Director, Naval Research Lab., Washington, D.C.	
Code 700, Metallurgy Division	1
Code 186, Tech. Lib.	1
Chief, Bureau of Yards and Docks, Washington, D.C.	1
Commanding Officer, U.S. Naval Ordnance Test Station, Inyokern, Calif	1
Commanding Officer, Naval Air Materiel Center, Philadelphia, Pa.	1
Commanding Officer, Watertown Arsenal, Massachusetts	1
Commanding Officer, Frankford Arsenal, Philadelphia, Pa.	1
Office of Chief of Engineers, Dept. of Army, Washington, D.C.	1
U.S. Air Forces Research & Development Division, Washington, D.C.	1
U.S. Atomic Energy Commission, Division of Research, Washington, D.C.	1
National Bureau of Standards, Washington, D.C.	1
National Advisory Committee for Aeronautics, Washington, D.C.	1
Director, Office of Naval Research, Branch Office, 346 Broadway, New York 13, New York	1
Director, Office of Naval Research, Branch Office, 86 East Randolph St., Chicago 1, Illinois	1
Director, Office of Naval Research, Branch Office, 1000 Geary Street, San Francisco 9, California	2
Director, Office of Naval Research, Branch Office, 1030 East Green Street, Pasadena 1, California	1
Dr. Hoylande D. Young, Argonne Nat'l. Lab., Chicago, Illinois	1
B. H. Fry, U.S. Atomic Energy Commission, 1901 Constitution Ave., N.W. Washington, D.C.	2
Carbide and Carbon Chemicals Div., Plant Records Dept., Central Files (K-25), Oak Ridge, Tennessee	1
Carbide and Carbon Chemicals Div., Central Reports & Information Office, (Y-12), Oak Ridge, Tennessee	1

Distribution List (Cont'd.)

	No. of Copies
Miss M. G. Freidank, General Electric Co., Technical Services Div., Tech. Information Group, Richland, Washington	1
Dr. F. H. Spedding, Iowa State College, Ames, Iowa	1
Document Librarian, Knolls Atomic Power Lab., Schenectady, New York . . .	1
Document Custodian, Los Alamos Scientific Lab., Los Alamos, New Mexico. .	1
Dr. J. J. Burgage, Mound Lab., U.S.Atomic Energy Comm., Miamisburg, Ohio.	1
Division of Technical Information and Declassification Service, N.Y., Operations Office, Ansonia Station, New York, N.Y.	1
Central Files, Oak Ridge Nat'l. Lab., Oak Ridge, Tennessee	1
Mr. Dale N. Evans, Sandia Corp. Sandia Base, Classified Document Division, Albuquerque, New Mexico	1
U.S. Atomic Energy Commission Library Branch, Technical Information, Oak Ridge, Tennessee	1
Dr. R. K. Wakerling, Radiation Laboratory, Univ. Of California, Berkeley, California	1
Librarian, Westinghouse Electric Corp., Atomic Power Div., Pittsburgh, Pa.	1
Asst. Naval Attache for Research, American Embassy Navy #100, New York. .	6
Dr. R. F. Mehl, Carnegie Institute of Technology, Pittsburgh, Pa.	1
Dr. Robert Maddin, Dept. of Mechanical Engineering, Johns Hopkins Univ. .	1
Dr. C. S. Smith, Inst. for Study of Metals, Chicago University	1
Dr. Henry Eyring, University of Utah, Salt Lake City, Utah	1
Dr. F. A. Biberstein, Jr., Catholic University, Washington, D.C.	1
Battelle Memorial Institute, Columbus, Ohio, Attn: Mr. H. C. Cross . . .	1
Mr. A. J. Sindeband, American Electro Metal Corp., Yonkers, N.Y.	1
Dr. R. L. Fullman, Research Lab., General Electric Co., Schenectady, N.Y.	1
Prof. Clark, Calif. Institute of Technology, Pasadena, California	1
Dr. Finn Jonasson, National Academy of Sciences, Washington, D.C.	1
Dr. J. S. Koehler, Carnegie Institute of Technology, Pittsburgh, Pa. . .	1
Prof. M. S. Sack, Cornell University, Ithica, New York	1
Prof. W. Prager, Brown University, Providence, Rhode Island	1
Dr. L. V. Griffis, Armour Research Foundation, Chicago, Illinois.	1
Dr. W. C. McGregor, Univ. of Pennsylvania, Philadelphia, Pa.	1
Dr. J. E. Dorn, University of California, Berkeley	1
University of California, Berkeley, Department of Engineering	3
Dr. Frederick Seitz, Physics Dept., University of Illinois, Urbana, Ill.	1
Dr. Thorton Read, Bell Telephone Laboratories, Murray Hill Laboratory, Murray Hill, New Jersey	1
Prof.. P. H. Beck, Dept. of Metallurgy, University of Illinois, Urbana, Ill.	1
Prof. W. M. Baldwin, Metals Research Lab., Case Inst. of Technology, Cleveland 6, Ohio	1
Prof. R. M. Brick, University of Pennsylvania, Philadelphia, Pa.	1
Prof. Thomas Read, Department of Metallurgy, Columbia Univ., New York, N.Y.	1
Office of Ordnance Research, Duke University, 2127 Myrtle Drive, Durham, North Carolina, Attn: Dr. A. G. Guy	1
H. Markus, Pitman Dunn Lab. Dept., Frankford Arsenal, Bridge and Tacony Sts., Philadelphia 37, Pa.	1
Prof. Charles S. Smith, Dept. of Physics, Case Institute of Technology, University Circle, Cleveland 6, Ohio.	1
Dr. R. D. Potter, 1407 Michelson Lab., China Lake, Inyokern, Calif. . . .	1

Distribution List (Cont'd.)

	No. of Copies
Dr. E. Orowan, Dept. of Mech. Eng., Mass. Inst. of Technology, Cambridge 39, Mass.	1
Brookhaven National Laboratory, Tech. Information Div., Attn: Research Lib., Upton, Long Island, New York	1
Dr. D. S. Billington, Oak Ridge National Laboratory, P. O. Box P, Oak Ridge, Tennessee, Attn: M. L. Bray, Supervisor, Central Files Dept., Information and Reports Division	1
John R. Cuthill, Enameled Metals Section, National Bureau of Standards, Washington 25, D.C.	1
Dr. N. J. Grant, Department of Metallurgy, Massachusetts Institute of Technology, Cambridge 39, Massachusetts	1
Prof. Charles Kittel, Physics Dept., Univ. of Calif., Berkeley, Calif.	1
Prof. M. Cohen, Mass. Institute of Technology, Cambridge 39, Mass.	1
Prof. A. Goldberg, U.S. Naval Post Graduate School, Monterey, Calif.	1
Dr. C. S. Barrett, Inst. for the Study of Metals, Chicago University, Chicago, Illinois	1
Dr. S. Smoluchowski, Carnegie Inst. of Tech. Schenley Park, Pittsburg 13, Pa.	1
Dr. S. Weissmann, Rutgers University, New Brunswick, New Jersey	1
Armed Services Technical Information Agency, Document Service Center Knott Building, Dayton 2, Ohio	5
Office of Technical Services, Dept. of Commerce, Washington 25, D.C.	1
E. H. Edwards, 4623 Elmwood Road, Richmond, California.	1
Dr. George C. Kuczynski, Metallurgy Dept., University of Notre Dame, South Bend, Indiana	1
Wright-Air Development Center, Wright-Patterson Air Force Base, Ohio	
Attn: Materials Lab. (WCRTL-6)	1
Aeronautical Res. Lab. (WCRRL)	1
Dr. B. D. Cullity, University of Notre Dame, South Bend, Indiana.	1
Office of the Commanding Officer, Department of the Army, Pentagon, Washington 25, D.C., Attn: ORDTB, E. L. Hollady	2